Do additive manufactured parts deserve better?

J.R.Yates[1], P. Efthymiadis[2], A. Antonysamy[2], C. Pinna[3], J.Tong[4]

Abstract

Additive manufacturing of metallic components is regarded as one of the more exciting developments in engineering. The combined attractions of near net shape, tailored composition and geometry optimisation have led to much interest in the various processes used and a drive to improve the mechanical properties to match those of wrought parts.

In this paper, we reflect on the apparent lack of ambition in optimising the structural integrity of parts made using these new manufacturing processes. The current research focus seems to be either on largely irrelevant static properties, or on quantifying the fatigue response in a way that would be familiar to engineers in the 19th Century.

Given the work on the role of microstructure and fatigue, which dates back to Ewing and Humphrey in 1903 reaching its zenith in the 1980s and 90s with Keith Miller in the vanguard, and recent developments in both imaging technologies and sophisticated numerical modelling, all the elements are in place for a much more rigorous, and ultimately more fruitful, approach to understand the structural integrity of additive manufactured components.

Keywords

Fatigue, additive manufacture, statistics of extreme values, Ti6Al4V

Introduction

Metal Additive Manufacturing (AM), considered by some [1] as part of the *third industrial revolution*, introduces significant freedom to design engineering components with improved functionality at potentially lower cost.

The attractions of AM include the possibility of fabricating components with complex geometry without dies or substantial machining, resulting in a reduction in lead-time, waste, and cost. Intricate features and internal structures are more readily formed, with the possibility of consolidating parts of previously complex assemblies. AM also offers a reduction in material consumption and waste generation, the rapid production of prototypes and reduced design iterations, resulting in shorter lead times and faster introduction of new products to market; truly the Holy Grail of modern engineering!

AM is commercially attractive for sectors like aviation, aerospace and others that involve low volume manufacturing, because of the reduced costs in producing bespoke parts. The buy-to-fly ratio of components produced via AM is roughly 1.5-5:1, with less material being wasted through machining, when compared to 10-20:1 for the normal ingot cast-roll-forging and machining route. This gives major advantages in terms of cost, especially when dealing with expensive reactive materials such as Ti, Co, Ni, and Cr based alloys.

[1] Sheffield Fracture Mechanics. Corresponding author: docjry@gmail.com
[2] GKN Aerospace
[3] University of Sheffield
[4] University of Portsmouth

Titanium alloys, particularly Grade 5 Ti6Al4V, are widely used as a lightweight material in modern aerospace structures, which need high structural efficiency, with high performance at moderate operating temperatures, as well as good fatigue and creep strength. Their increased usage in advanced commercial aircraft designs, with the Boeing 787 aircraft containing about 20% by weight of titanium alloy, means that reducing the manufacturing costs of Ti6Al4V component by AM techniques is of great interest to industry, and hence researchers.

A major concern in the use of titanium alloys in other fields, such as the automotive and chemical industries, comes from the high cost of conventional production methods and the challenges of ensuring acceptable levels of quality. If AM can overcome these hurdles, then the scope for widespread adoption of additive manufacturing is immense.

A number of challenges, however, are hampering the complete deployment and full adoption of the AM technology. The main issues include the complexity of manufacturing process controls including the need for high vacuum, the questionable applicability of conventional non-destructive inspection methods, the lack of industry standards, the inherent process-related defects of AM materials and components, limited opportunity to modify the microstructures after AM processing since products are fabricated to near-net shapes, low deposition rates and build volumes, and high production costs.

Furthermore, there is a lack of field experience with AM components, particularly in safety critical industrial applications. In these industries, service life and durability are dependent on the fatigue properties of component, and those made by AM usually display inferior performance to conventional wrought and machined parts.

Whilst standards agencies, for example [2-4], are heavily involved in devising procedures for approval, qualification and certification of manufacturers of additive materials, there is a tendency to base these on testing methods developed for conventional cast, wrought or machined components. As Seifi *et al*. [5] point out:

> "*For AM, being a relatively new manufacturing technology, the specific testing procedures still need to be developed, reflecting the unique nature of AM material systems including anisotropy, inherent material anomalies, location-specific properties, residual stresses, etc*."

They also note that the FAA Advisory Circular 33.70-1 [6] advises that a probabilitistic approach is one of two elements required for a damage tolerance assessment for aircraft engine life limited parts. This leads to their recommendation that:

> *"…the appropriate characterization of material anomalies is needed, in addition to conventional fatigue and fracture properties of substrate materials. Such characterization should focus on developing the size distribution and frequency of occurrence of material anomalies. This information can be used to define an exceedance curve for a given class of material defects, which is the key input into probabilistic fracture mechanics based assessment,…"*

There is much published research, and many reviews, on the fatigue behaviour of test specimens manufactured in a wide variety of metallic systems by a multitude of AM processes. This paper is not intended to add to that corpus; instead we shall reflect on the inherent nature of AM components and materials and offer thoughts on other ways of tackling these problems based on the vast body of scientifically useful work that exists on the physical mechanisms of fatigue.

Historical perspective

Man has been making metallic objects for over 5000 years. A decorative copper frog from 3200BC is thought to be one of the oldest known cast metallic objects; shortly followed by functional bronze tools and weapons, clearly showing the early metallurgical recognition that alloying could enhance mechanical properties. Hot forging appears to predate casting, and again seems to have started in the region between the Tigris and Euphrates rivers. Gold was almost certainly the first metal to have been worked into creative objects, and as extraction metallurgy developed copper and copper alloys found more practical use. The development of cast iron, wrought iron and eventually steel in all its forms is well documented, although interestingly iron was routinely cast in China around 1800 years before Europe.

The Industrial Revolution in Northern Europe in the late 18th Century introduced machine tools and mechanisation which imposed new operating conditions on metallic components. This led to the new phenomenon of failure after a period of usage rather than failure on first loading. The first recorded account of this is usually attributed to a German mining engineer, Wilhelm Albert, who in 1829 observed and reported on the failure of mine hoist chains made from iron. He constructed a test machine and found that failures were associated with the magnitude of the loads and the number of repetitions, and were not due to accidental overloads [7]. Jean-Victor Poncelet, the French mathematician, academic and engineer is credited with the first use of the term 'fatigue' in his 1839 book on mechanics [8]. So, we might reasonably suppose that the concept of failure of machinery after a period of use was well known to engineers in the early 19th Century.

The rapid expansion of railways in the first half of the 19th Century gave engineers plenty of opportunity to deal with fatigue failures, and it was a topic of much interest to the scientists and mathematicians of the time. Following the disaster at Versailles in 1842 William Rankine [9] and Joseph Glynn [10] recognised that fatigue was a process of crack growth and not a result of the then commonly held view of a metal becoming crystalline and hence brittle. Rankine identified that the shape of design features was critical and Glynn, very perceptively, stated that the fatigue failure started with the first journey.

The physical mechanisms of fatigue became clear in the early part of the 20th Century with Sir James Ewing and Joseph Humfrey [11] identifying in 1903 that microcracks were formed by slip within grains. By 1924 Herbert Gough [12] had clarified the role of slip systems on the formation of fatigue cracks, and by the early 1960s the distinction was made between Stage I and Stage II cracks by Peter Forsyth [13]. The absolutely critical point was that growth of Stage I cracks was governed by the non-continuum, highly oriented slip within individual grains and perturbed by grain boundaries, and Stage II cracks were describable by conventional continuum mechanics.

Meanwhile, in parallel and seemingly oblivious to the scientific understanding of fatigue cracking, the tools used by industry to deal with fatigue and durability were entirely based on empirical correlations between macroscopic parameters such as the cyclic range of nominal stress, local strain, or stress intensity factor and observable events such as failure of a test specimen, appearance of a small crack, or increment of crack extension per loading cycle. Depending on one's preference, S-N or Wohler curves, strain life or the Coffin-Manson equation, or the Paris Law were simple to understand, attractive in the ease with which experimental data may be obtained, straightforward to encode in software and hence ultimately commercially successful. The enthusiasm with which modifications, enhancements, refinements and complexity were added to these empirical methods served only to reinforce their position as the *de facto* tools for the fatigue analysis of metallic components.

Characteristic defects in AM materials and components

The industrial process of additive manufacturing of metallic parts is in its infancy; it is only 40 years since the first trials of selective laser sintering. Whilst the intention is to produce components of high integrity and durability, comparable with conventionally made parts, the current reality is that the fatigue performance tends to be inferior.

There are many reviews of this subject; for the key issues see Molaei and Fatemi [14]. It is well established that AM introduces material anomalies in the form of gas porosity, lack of fusion, inclusions, micro-cracking as well as surface defects. In addition, microstructures often differ from wrought products for the same alloy system, and there may be internal residual stresses that again differ from those parts made by other manufacturing routes. Additional processes may be introduced, at additional cost, to deal with some of these limitations. Hot Isostatic Pressing (HIP) is used, in an effort to de-risk AM parts at critical, highly loaded locations, to reduce the embedded residual stresses and porosity within the part.

The vast majority of published studies on the fatigue of AM material tend to present the finding in terms of conventional stress-life, strain-life or long crack propagation data. However, a question one might reasonably ask is:

> '*why should empirical correlations of fatigue devised for metals with several thousand years of metallurgical development be suitable for these new, highly defective, microstructures?*'

This question is more pertinent when one considers that the fundamental physical mechanisms of fatigue are well known and there exist a wide range of models to describe them. Surely, a mechanistic approach to predict the fatigue behaviour of AM materials would be more fruitful since they are, in essence, conventional metallic alloys that are riddled with defects.

Short cracks in fatigue

The early formation of fatigue cracks and their subsequent crack growth to the point where conventional continuum mechanics can be appropriately utilised are the two phases of crack development which are most influenced by the microstructure [15]. The substantial scatter commonly observed in experimental results [16], and hence the high uncertainty in fatigue lifetime predictions, arise from the complex interplay between the varying local strains within the microstructure, the orientation and distribution of microstructural features, and the size and spatial distribution of different types of defects. All of this is well known, and one may look at the modelling work in a range of polycrystalline metals, such as aluminium alloys [17-19], titanium [20, 21], stainless steels [22] and nickel-based super-alloys [19, 23], to appreciate the breadth of knowledge and expertise developed over the last decades.

The physical mechanisms that lead to fatigue crack formation, or initiation, are highly localised. For instance, dislocation dipoles, interface decohesion, triple points, second-phase particles can lead to strain localisation followed by crack nucleation [16, 20, 21, 24]. The propagation of a short crack depends on the resistance and strength of the matrix. Depending on the location of the crack, whether it is close or not to a grain boundary, crack acceleration or deceleration can occur [25]. Particles embedded in the matrix may facilitate further propagation if the particles break in a brittle manner or retard crack growth if the particles break in a ductile manner. As a result of the multiplicity of competing mechanisms, the process of localised damage nucleation and short crack growth can be very complex even in conventional alloys. In AM materials one must also add the larger scale defects of gas porosity, lack of fusion and so on. A reliable physically-based fatigue model must therefore capture the effects of the various mechanisms that may contribute to the formation of a fatigue crack at the dominant scale of the microstructure.

Many computational approaches, such as crystal plasticity finite element (CPFE) modelling, molecular dynamics, discrete dislocation dynamics and conventional continuum-based finite element (FE) modelling have been utilised with different criteria to model early fatigue crack growth [16, 17-23, 26, 27]. CPFE modelling has been shown to have merit [28] and several criteria have been proposed ranging from energy based approaches, for example Cheong *et al*. [26] and critical accumulated slip, such as that of Manonukul and Dunne [15] to combinations of slip and energy terms, including Shenoy *et al*. [16], Bozek *et al*., Hochhalter *et al*. [17-19], and Efthymiadis *et al*. [29].

In many ways, the fatigue of AM materials ought to be rather more straightforward; the dominant features of porosity and lack of fusion are of much larger length scale than the grain scale features of high integrity wrought metals. Until the manufacturing processes become sufficiently competent to produce components with defects at the micron scale rather than those two orders of magnitude larger, there ought to be scope for a mesoscale approach to fatigue.

The stochastic nature of the size and spatial distribution of critical defects in AM materials suggests that probabilistic models may be more suitable than strictly deterministic approaches. Statistical methods based on Monte Carlo simulations or the concept of statistics of extreme values have shown some promising results for the quality control and fatigue strength prediction of AM materials, see for example, [30, 31].

In the remainder of this paper we shall adapt a statistical model originally developed for fatigue from inclusions in ultraclean bearing steels [32] to revisit the work on porosity in an electron beam melting (EBM) titanium alloy by Tammas-Williams *et al*. [33].

Size and spatial distribution of porosity

Tammas-Williams *et al*. [32] have reported on a thorough study of the size distribution of pores in EBM Ti6Al4V using time lapse X ray computed tomography and linked this with the fatigue performance in terms of stress-lifetime response. They were able to link the final failure of test specimens with the size of the pores occurring at or near the surface. They found that it was not necessarily the largest pore that led to failure, but a combination of pore size, proximity to the surface and the local stress that was important. This mirrored the findings of Yates *et al*. [32] who studied the size distribution of oxide inclusions in clean steels. They concluded that it was the combined probability of finding a large, but not necessarily the largest, inclusion at a high, but not necessarily the highest, stress site that gave the greatest probability of fatigue failure.

Yates *et al*. used a Generalised Pareto Distribution (GPD) to characterise the inclusion size distribution and we have adopted the same to review the pore size distribution measured by Tammas-Williams *et al*. There are three terms to the GPD: a threshold parameter u, a shape parameter ξ ($-\infty<\xi<\infty$), and a scale parameter $\sigma'>0$. Suppose *y* is the size of a pore greater than the threshold *u*, then the cumulative distribution function, *F(y)*, for a pore no larger than *y*, given that it exceeds *u*, is given approximately by the GPD function:

$$F(y) = 1 - \left(1 + \frac{\xi(y-u)}{\sigma'}\right)^{-1/\xi} \qquad \text{(Eq. 1)}$$

By considering only values that are larger, or indeed smaller, than a given threshold, known as exceedences, then the extreme tails of a distribution may be modelled independently of the data around the mean or mode. The attraction of the Generalised Pareto Distribution is the ability to model the tail of a distribution without restricting the shape or form of that tail. The shape parameter ξ in Eq. 1, when fitted to data, describes the limiting distribution of the exceedence data. For example, if $\xi=0$ then the tail decreases exponentially, as in a normal distribution; if $\xi<0$ then the tail is finite; and if $\xi>0$ then the tail decreases as a polynomial, as in Student's *t*, see Figure 1.

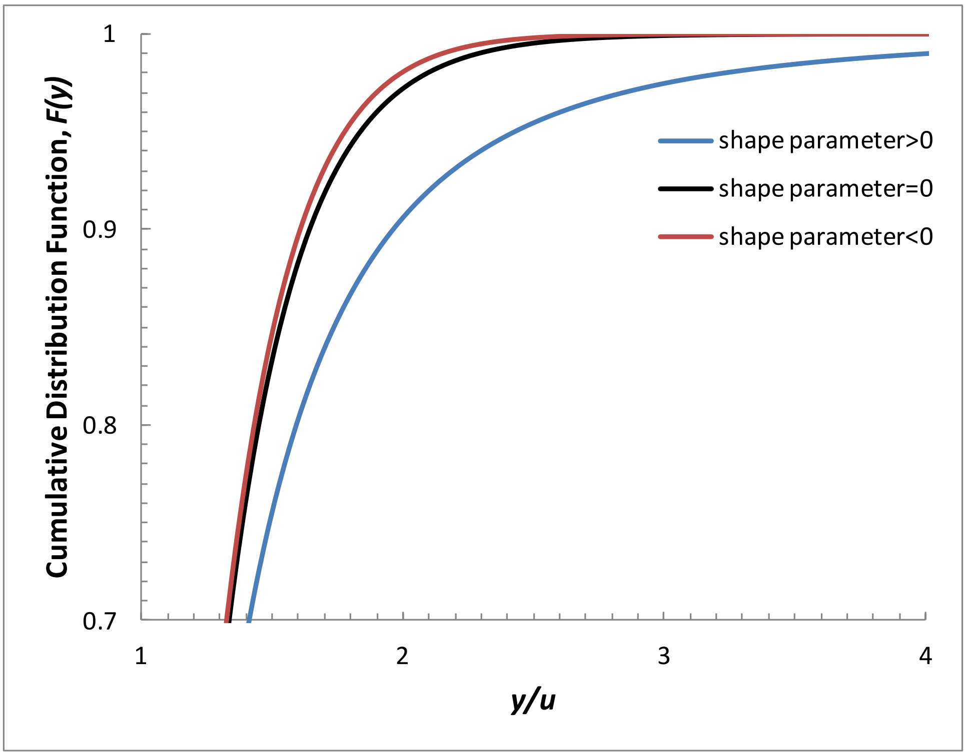

Figure 1. Cumulative Distribution Function for the Generalised Pareto Distribution for u=25 μm, σ' = 7, and shape parameters, ξ =0.33, 0 and -0.05

It is probably of more interest to consider the complementary cumulative distribution function, *P(y>u)*; the probability of observing a test statistic, *y*, at least as extreme as the one observed, where

$$P(y > u) = 1 - F(y) \qquad \text{(Eq. 2)}$$

Plotting this, as a logarithm, with positive, negative and zero shape parameters in Figure 2 clearly shows the different forms of the test statistics in the tails of the distributions. Of particular note is the finite limit when $\xi<0$. Furthermore, it can be seen that there is a higher probability of observing a test statistic, *y*, at least as extreme as the one observed for $\xi>0$, such as in Student's *t*, than $\xi=0$, as seen in the normal distribution.

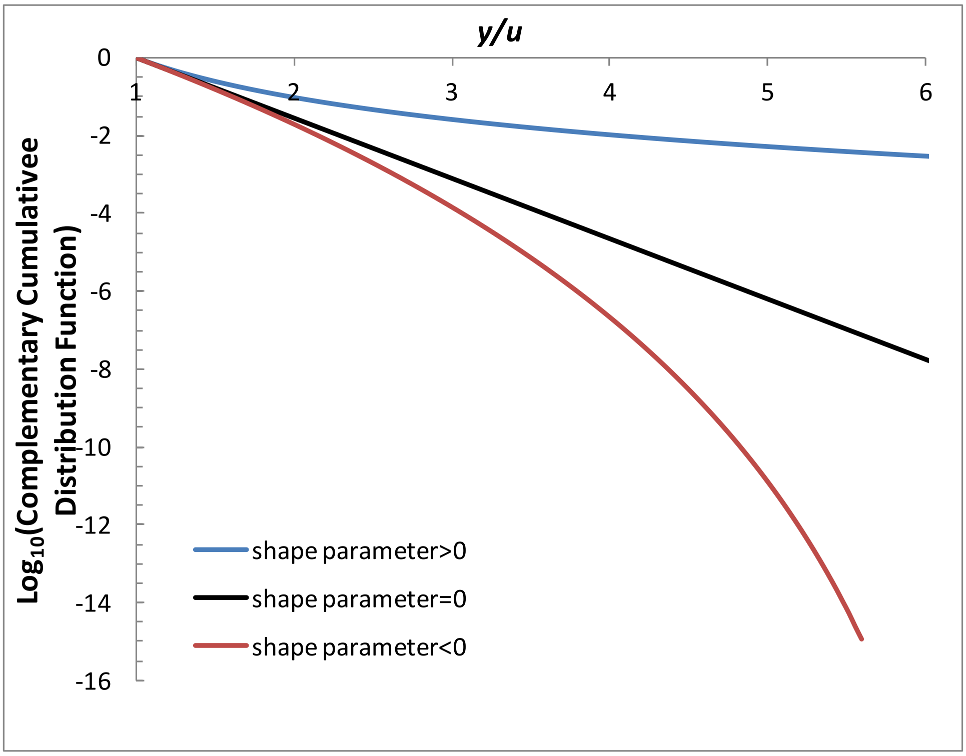

Figure 2. Complementary Cumulative Distribution Function for the Generalised Pareto Distribution u=25 μm, σ' = 7, and shape parameters, ξ =0.33, 0 and -0.05

In problems of fatigue where the formation of the critical crack is governed by some microstructural anomaly, such as an inclusion in clean steels, then it is the size of the large but rare inclusions that dominate the durability. The GPD is a useful tool to characterising the statistics of extreme values for such cases.

Considering the pore size data in Tammas-Williams *et al.*, we are able to fit the GPD with the parameters u=25 μm, ξ = 0.33 , σ' = 7, where the threshold is consistent with the pore detection limit of 26 μm quoted, as shown in Figure 3.

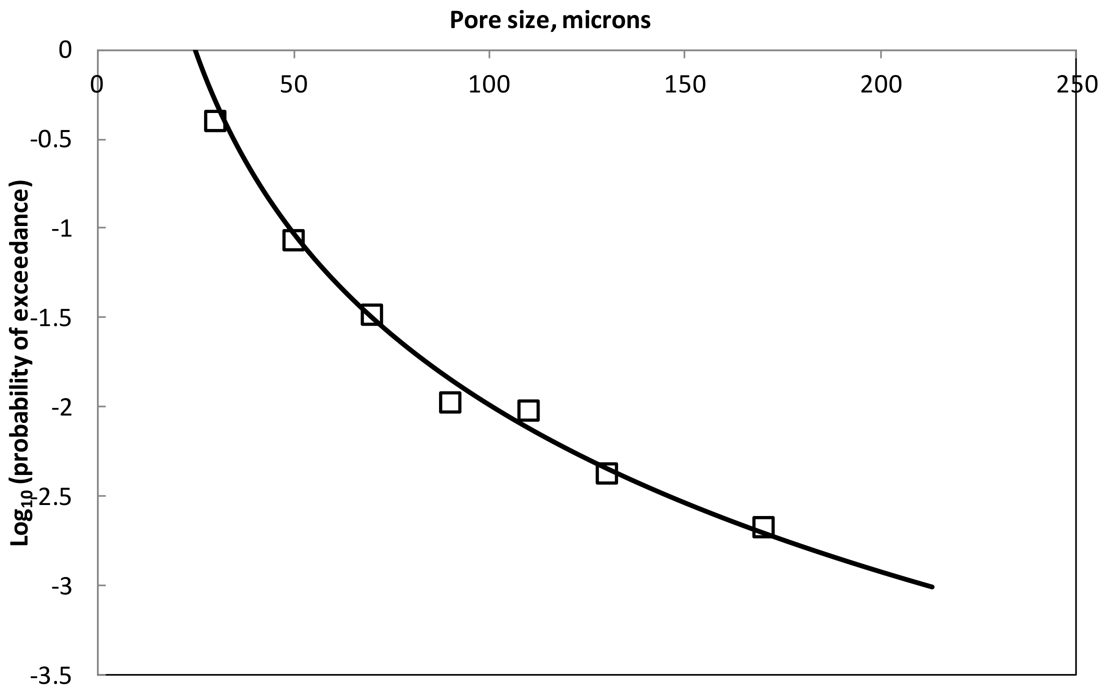

Figure 3. Generalised Pareto Distribution fitted to Tammas Williams *et al.* [33] pore size data.

Estimating the fatigue life

We shall start from the premise that the fatigue life of small, highly stressed test specimens of Ti6Al4V made by additive manufacturing, at least those made by the electron beam melting process used by Tammas-Williams *et al.*, is dominated by the size of large pores located near the surface, and not by the finer scale microstructure, nor by the presence of small microcracks or inclusions. In this case, if we have some information about the probability of finding pores that exceed a certain size, and information about the growth rate of fatigue cracks in essentially a defect free microstructure, we ought to be able to construct stress-life curves for different pore sizes.

Tammas-Williams *et al.* used a relatively small diameter cylindrical specimen of 4.5mm diameter and, from their micrographs, one can observe that final fracture occurred when a fatigue crack had grown across about half the section. To estimate the lifetime we need some fatigue information appropriate to the size of the flaws in these specimens. We therefore need a crack propagation equation to cover the range of 0.1mm, typical of the larger pores, to around 2mm, which would break a 4.5mm diameter specimen. Fortunately these sizes are larger than the microstructurally dominated short fatigue crack growth in Ti alloys as this would pose considerable difficulties in acquiring data for the particular microstructure as manufactured. The crack sizes we are interested in fall into the range of physically short cracks; those cracks that are small relative to their plastic zones and have not developed the full contact or closure in their wake as expected for a long crack from a traditional fracture mechanics test specimen. Physically short crack growth data for Ti alloys is not readily available but we are fortunate that Zhai *et al*. [34] have published such data for Ti6Al4V.

The data generated and published by Zhai *et al*. provides evidence of the microstructurally short, physically short and conventional long crack propagation of both a laser deposited material in two orientations and the monolithic substrate upon which the AM material has been fabricated. Interestingly, the differences in growth rates between the monolithic and the AM fabricated material, and the differences between the orientations are relatively small, and certainly smaller than the variations in published growth rates documented by different researchers.

In postulating that the fatigue of AM materials is governed by a combination of the distribution of defects and the propagation of cracks through a homogeneous microstructure, we have taken the physically short crack growth data for the substrate material in the vertical direction at R=0.1

$$\frac{da}{dN} = 1.8\times10^{-13}\Delta K^{4.7} \text{ m/cycle for } \Delta K \text{ in MPa}\sqrt{\text{m}} \qquad \text{(Eq. 3)}$$

The fatigue lifetime is simply the integration of Eq. 3 between the initial defect size and a final crack length. By assuming a constant value, 0.65, for the geometry correction term for the stress intensity factor of a surface, or near surface flaw, and the usual $\sqrt{area}$ argument for defect size [35], we obtain an analytical solution for the integration. The lifetime is relatively insensitive to the value of the final crack length; we have chosen 2mm.

The GPD allows us to estimate the likelihood of pores found by Tammas-Williams *et al*. larger than a given size, given that they exceed the threshold size of 25 microns, see Table 1.

Probability of pore larger than *y* occurring	Pore dimension, *y*, microns.
0.1%	200
1%	100
2.5%	75
10%	50

Table 1. Probability matrix of pores with a given size

Using each of these pore sizes as initial defect sizes in the crack growth calculation yields the stress-life curves, shown in Fig. 4. Also included on the graph, courtesy of Tammas-Williams *et al*., are the initiating defect sizes measured for each of their test specimens.

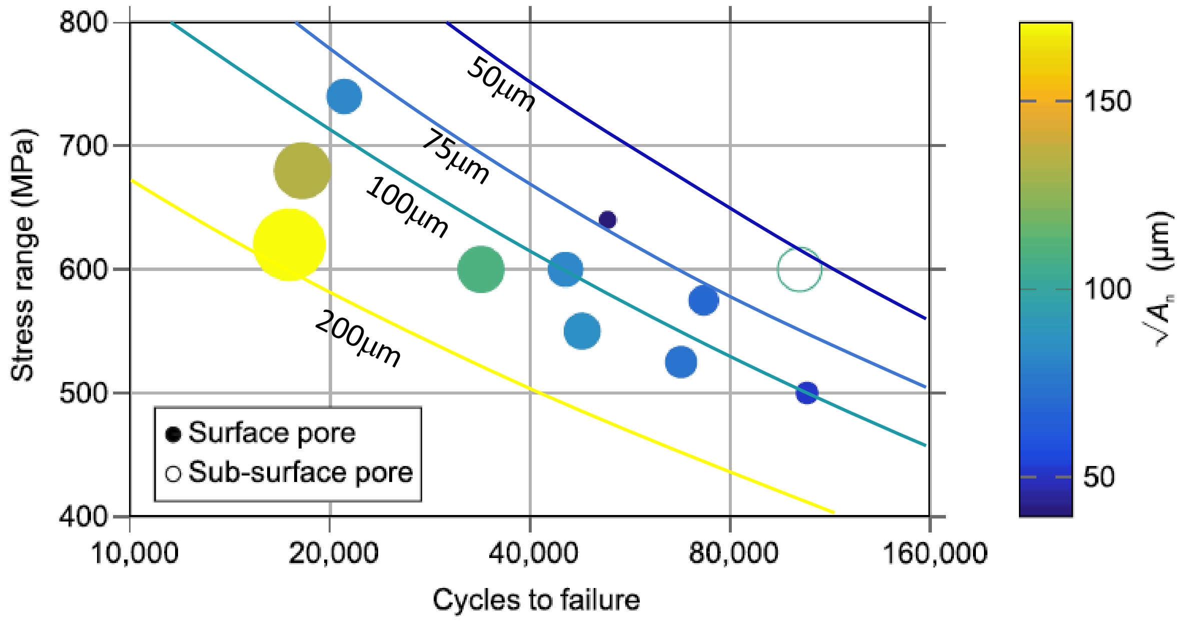

Figure 4. Experimental results from Tammas-Williams *et al*. [33] with calculated stress-life curves. The data points and curves are colour coded with the same scheme.

It can be seen from Figure 4 that the agreement between the lifetimes estimated by our simple analysis and the experimentally measured fatigue lives and their associated initiating pore dimensions is remarkably good. Whilst the fatigue model developed herein is very simple, it leads to a number of intriguing observations and further questions.

One important outcome is the ability to associate the probability of finding large pores with the spread in observed fatigue lives; rare but large pores result in short lifetimes. An interesting aspect of this is the positive value of ξ which suggests, at least mathematically, that there is no upper limit to the maximum size of pore in these materials. Physically, this is not unreasonable as Tammas-Williams *et al*. observed crack initiation from conjoined pores. If large pores are fairly common, and there are observations of conjoined pairs, then one must accept that there is a finite probability, in a large volume of material, of linking three, five, or even ten pores to create a rare superflaw. Since the fatigue lifetime declines with initial pore size, the consequence of the possibility of such a large pore is the possibility of a near zero fatigue life. In practice, one might expect that inspection of parts would identify those with unacceptably large defects, but clearly, reducing the population of large pores would, at the very least, reduce the component reject rate.

Given the sensitivity of the fatigue life to pore size in this type of additive manufactured material, the GPD may be a useful tool with which to examine the influence of process variables on porosity; potentially identifying process routes that yield a negative shape factor to the tail of the defect size distribution, and hence a finite limit to the size of initiating defects.

A potentially useful extension of this simple approach would be to apply the analysis developed by Yates *et al*. [32], and supported by the work of Tammas-Williams *et al*. [33], to create a fatigue model for additive manufactured components of complex shape and loading. Since the fatigue failure is a combination of the likelihood of finding a large, but not necessarily the largest flaw in a location with a high, but not necessarily the highest, stress a probabilistic approach might prove to be fruitful.

The wide variety of additive manufacturing processes in use for metallic components gives rise to a wide variety of metallurgical and structural defects. We may have been fortunate to have chosen a case in which one type of defect appears to dominate the fatigue behaviour. Nevertheless, the approach illustrated ought to be widely applicable to other alloy systems and other manufacturing methods, provided one can identify and characterise the dominant crack formation mechanism. In cases where two or more mechanisms are in competition, each should be treated separately and then combined in, perhaps, a probabilistic model, of which Monte Carlo methods can be useful.

Outlook

In contrast to conventional manufacturing methods, there are limited options to modify the microstructure or surface finish after AM processing without additional cost, as products are fabricated to near-net shape. The unidirectional heat flow during direct printing AM process often results in the formation of elongated grains parallel to the build direction, resulting in anisotropic microstructure and mechanical properties of parts along and perpendicular to the build direction, which are further enhanced by the differences in the build thermal history [36,37]. The impact of process design parameters on physical and mechanical properties of AM materials and their components is not well understood, particularly under service loading conditions where variable loads and other extreme conditions may apply. A fundamental understanding of the influence of process parameters on the microstructural evolution, subsequently mechanical behaviour of AM materials and parts is therefore of vital importance to predicting the performance of AM parts in service [38,39].

For applications in safety-critical industries, service life and durability are dependent on the fatigue properties of materials and components made through AM routes [40]. AM produces process-dependent microstructures and typical features related to solidification of the surface layers [41]. The latter appear to be the immediate relevant issue affecting fatigue life in AM materials [33], where as-built specimens exhibit a dramatic reduction of 40–50% of fatigue strength compared with those machined [42-45]. Significant surface roughness acts as multiple stress concentration sites, whilst tensile residual stresses, subsurface pores and defects may promote crack initiation [46], although the precise mechanisms are yet to be elucidated. AM processes can introduce a number of material anomalies such as gas porosity, lack of fusion, inclusions and micro-cracking [47-49], and each class of anomalies has a different formation mechanism, and its impact on fatigue and fatigue crack growth also differs. It is therefore extremely important to understand the effects of each type of material anomaly from AM processes on the structural integrity of the specimens and the components in the loading regime of interest, to characterise the most influential anomaly types and to quantify their impact on the resulting part performance for safety-critical AM applications.

Tools and techniques developed over recent years in both experimentation and multiscale modelling are suitable for fundamental studies of fatigue behaviour of AM materials and components. Such work would contribute to meeting the needs of the standards agencies [2-4] through a generic roadmap for smart AM designs, enabling optimum performance and assured mechanical integrity of AM parts with knowledge of the statistical influences highlighted by the FAA [6]. Such an assessment framework may be developed through realistic process and post process simulations, deformation and damage constitutive modelling, together with structural integrity assessment routines.

Conclusions

The structural integrity assessment of additive manufactured metallic components must include both knowledge of the statistical variation of the material microstructure and a fatigue lifing methodology firmly rooted in the physical processes involved.

The pore size distribution of an electron beam melted titanium alloy has been shown to be well characterised by a Generalised Pareto Distribution. The GPD provides useful insight into the likelihood of finding large pores in additive manufactured metals.

Fatigue lifetime predictions made by integrating a straightforward physically short crack growth law starting from the pore sizes observed, and informed by the likelihood of such porosity from the GPD, show good agreement with experimentally measured fatigue lifetimes.

The discussion and analysis presented in this paper suggests that all the elements needed to predict the fatigue response of additive manufactured components already exist in the open literature. It is the complexity of the microstructure of AM components that needs attention; the way pores, flaws, grains and their textures, inclusions, residual stresses and structural stresses interact and the length scales that dominate those interactions that must be resolved.

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